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Loss of Fracture Resistance in High Carbon Steel Due to Non-Martensitic Phases
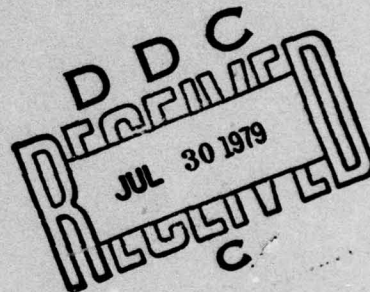
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Microstructure	Martensite	

20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

Attempts to control the dynamic compression failure modes of tungsten fiber reinforced steel matrices included changing the matrix material from mild steel to a hypereutectoid steel. However, the thermo-mechanical processes used to fabricate such metal matrix composite materials produce mixed microstructures, and it was not known to what extent the nonmartensitic transformation products influence mechanical behavior. This work examines the effects of isothermally transformed pearlite, bainite, and cementite on the strength, ductility, and plane-strain fracture toughness

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20. Abstract (Continued)

of a hypereutectoid silico-manganese steel that typifies the compositions being evaluated for a class of metal matrix composite systems. The small-grained, quenched-and-tempered martensite was the toughest material (60 MPa√m). Increasing the pearlite content gradually decreased the fracture toughness. However, increasing the bainite content by even slight amounts produced a catastrophic decrease in fracture toughness to half that value, which was found to be nearly the minimum K_{IC} determined from low temperature (-54C) tests, i.e., 28 MPa√m. This result dramatizes the importance of careful heat treatment when using such materials in engineering structures.

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CONTENTS

LOSS OF FRACTURE RESISTANCE IN HIGH CARBON STEEL
DUE TO NON—MARTENSITIC PHASES

INTRODUCTION

Current research on metal matrix composites in this laboratory has focused on the induction and control of specific compressive failure modes under high strain rate loading. In the case of tungsten fiber reinforced steel matrices, the mechanical properties of the matrix material dominated the failure mode and suggested that a shift from mild steel to a hypereutectoid composition might be desirable. However, the thermo-mechanical processes used to fabricate the composite materials produced mixed microstructures, and it was not known to what extent the non—martensitic transformation products influenced the strength and toughness of the steels.

Several investigators have examined the modes of tensile fracture in isothermally transformed hypereutectoid steels and shown them to result from shear cracking of pearlite colonies (1), intergranular cracking of cementite films along prior austenite grain boundaries (2) or intergranular cracking along bainitic ferrite grain boundaries (3). This latter case is attributed to cementite being the primarily nucleated phase in the isothermal transformation of high carbon steel from austenite to upper bainite. Numerous regions of high carbon austenite are entrapped by the newly formed bainitic ferrite laths and eventually transform to carbides along the ferrite grain boundaries. Such a process produces significant lengths of nearly continuous low energy fracture paths.

None of the previous studies has sought to determine quantitative effects of selected amounts of non—martensitic isothermal transformation products on the plane-strain fracture toughness of a material. The work reported here examined the effects of isothermally transformed pearlite, bainite, and cementite on the strength, ductility, and plane-strain fracture toughness of a hypereutectoid silico-manganese steel that typifies the compositions being evaluated for a class of metal matrix composite systems.

EXPERIMENTAL PROCEDURES AND RESULTS

Because of the difficulties to be expected in unraveling the kinetics of mixed microstructures produced by continuous cooling of

1

steels, this study was limited to simulative isothermal transformation experiments wherein the final quenches produced materials with predetermined percentages of pearlite or bainite in martensite matrices. In addition, some specimens were subjected to extremely high austenitizing temperatures to enhance grain growth and then isothermally treated to precipitate a grain boundary carbide (4).

The material used in this study was a hypereutectoid silico-manganese steel, designated HF-1, which was commercially available as 4 x 4 RCS (10 cm x 10 cm round-corner-square) bar stock. Chemical analysis of the as-received material disclosed the following elemental composition:

C	Mn	Si	P	S	Ni	Cu	Cr	V	Mo	Aℓ
1.08	1.88	0.90	0.002	0.014	0.1-0.2	0.06	0.1-0.2	<0.01	0.03	<0.01

Prior to committing substantial amounts of material to specific heat treatments, it seemed prudent to confirm the validity of the HF-1 steel T-T-T diagram at least in the regions of interest because the available diagram supplied by the manufacturer seemed to be of a preliminary nature for a material undergoing development. Moreover, such a diagram shows only start and finish times for reactions and is of somewhat limited value for intermediate percentages of the various phases. To resolve this problem, one hundred test specimens were prepared. Ninety were chips 30 mm x 30 mm x 3 mm, while ten were small plates 75 mm x 75 mm x 25 mm. The T-T-T data were generated by austenitizing ten chips at a time at 829C for one hour and then quickly transferring them to salt baths held at the desired transformation temperature. Various degrees of isothermal transformation were obtained by holding the chips at temperature for different times ranging from two seconds to 1000 seconds and then water quenching. This procedure was followed for nine equally spaced (~28C) temperature intervals over the range 454C to 677C, a range which encompasses both the pearlite and bainite noses of the T-T-T diagram. The as-quenched materials were polished and etched with Picral. An image analyzing computer was then used to evaluate the percentages of transformation products, and selected samples were evaluated using conventional point count techniques to double-check. The methods generally agreed within a few percent and the experimental T-T-T data are summarized in Figure 1 and compared to the manufacturer's data. It may be noted that in our experiments the pearlite reaction went faster than expected while, conversely, the bainite reaction went slower. The start times are in good agreement, while the finish times are reasonable. To verify that the same degrees of transformation could be obtained in moderately thick specimens which are oil quenched, the plates were tested by austenitizing for one hour at 829C and then isothermally transforming five each at 616C and 482C for selected times between five and 1000 seconds and then oil quenching. The microstructures agreed with those obtained by water quenching the small chips.
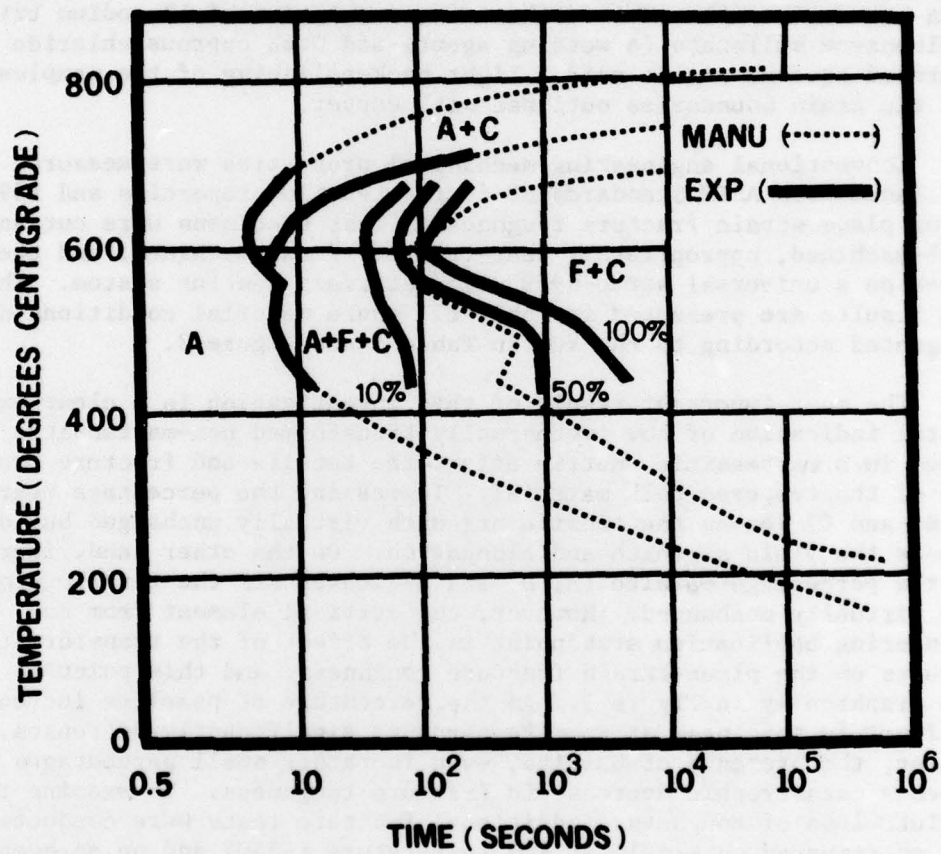
FIGURE 1. ISOTHERMAL TRANSFORMATION DIAGRAM FOR HF-1 STEEL

The heat treatment schedules used to produce the microstructures desired for this study are presented in Table I, and typical microstructures are illustrated in Figure 2. As-quenched samples are shown which represent martensite (A), pearlite in martensite (B and C), and bainite in martensite (D and E). Also shown are the tempered martensites that were isothermally treated to precipitate proeutectoid grain boundary carbides (F, G, and H). The larger grain sizes due to higher austenitization temperatures are also clear (G and H). As-quenched materials were simply polished and etched with Picral. However, the tempered martensites required a special etchant to resolve the prior austenite grain boundaries (5). This etchant was a solution of 1% sodium tridecylbenzene sulfonate (a wetting agent) and 0.6% cuprous chloride in saturated aqueous picric acid. Light back-polishing of the samples left the grain boundaries outlined with copper.

Conventional engineering mechanical properties were measured in accordance with ASTM Standards E8-69 for tensile properties and E399-74 for plane strain fracture toughness. Test specimens were cut and rough-machined, appropriately heat-treated, final-machined, and then tested on a universal servo-hydraulic materials testing system. The test results are presented in Table II, where material conditions are designated according to the key in Table I and Figure 2.

The most important result of this investigation is a clear experimental indication of how isothermally transformed non-martensitic phases in a martensitic matrix affect the tensile and fracture properties of the tempered bulk material. Increasing the percentage pearlite (A, B, and C) leaves the tensile strength virtually unchanged but decreases the yield strength and elongation. On the other hand, increasing the percentage bainite (A, D, and E) leaves all the tensile properties virtually unchanged. However, the critical element from an engineering application standpoint is the effect of the transformation products on the plane-strain fracture toughness, and this point is best made graphically in Figure 3. As the percentage of pearlite increases, the fracture toughness at room temperature significantly decreases. However, the presence of bainite, even in rather small percentages causes a catastrophic decrease in fracture toughness. To examine the absolute loss of toughness, additional fracture tests were conducted both on tempered materials at low temperature (-54C) and on as-quenched, i.e., untempered materials at room temperature. It is clear that in the process of transforming to upper bainite, the fracture toughness of this material decreases to near its minimum value.

The three remaining conditions (F, G, and H) do not admit the planned simple comparisons on the basis of grain size because the resultant tensile properties were difficult to control and could not be conveniently normalized. However, the generally low values of fracture toughness can be accounted for by observing that the proeutectoid carbides precipitated into nearly continuous networks throughout the grain boundaries and provided extensive paths for easy fracture.

4

TABLE I. FINAL HEAT TREATMENTS

Designation	Schedule	Microstructure*
A	829C, 1½ hr; OQ; AC, 1 hr; 343C, 1½ hr; 616C, 1½ hr; AC	100% Martensite
B	829C, 1½ hr; 616C, 30 sec; OQ; AC, 1 hr; 343C, 1½ hr; 616C, 1½ hr; AC	10% Pearlite – 90% Martensite
C	829C, 1½hr; 616C, 60 sec; OQ; AC, 1 hr; 343C, 1½ hr; 616C, 1½ hr, AC	50% Pearlite – 50% Martensite
D	829C, 1½ hr; 482C, 60 sec; OQ; AC, 1 hr; 343C, 1½ hr; 616C, 1½ hr; AC	10% Bainite – 90% Martensite
E	829C, 1½ hr; 482C, 600 sec; OQ; AC, 1 hr; 343C, 1½ hr; 616C, 1½ hr; AC	50% Bainite – 50% Martensite
F	774C, 1 hr; OQ; 593C, 1 hr; AC	100% Martensite Grain Boundary Network
G	927C, 1 hr; 760C, 1 hr; OQ; 482, 1 hr; AC	100% Martensite Grain Boundary Network
H	982C, 1 hr; 760C, 1 hr; OQ; 649C, 1 hr; AC	100% Martensite Grain Boundary Network

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*   Final microstructures of mechanical test specimens were fully tempered. Martensites generally had some small amount of retained austenite. The grain boundary network in F, G, and H refers to a more or less continuous network of proeutectoid carbides along the prior austenite grain boundaries.
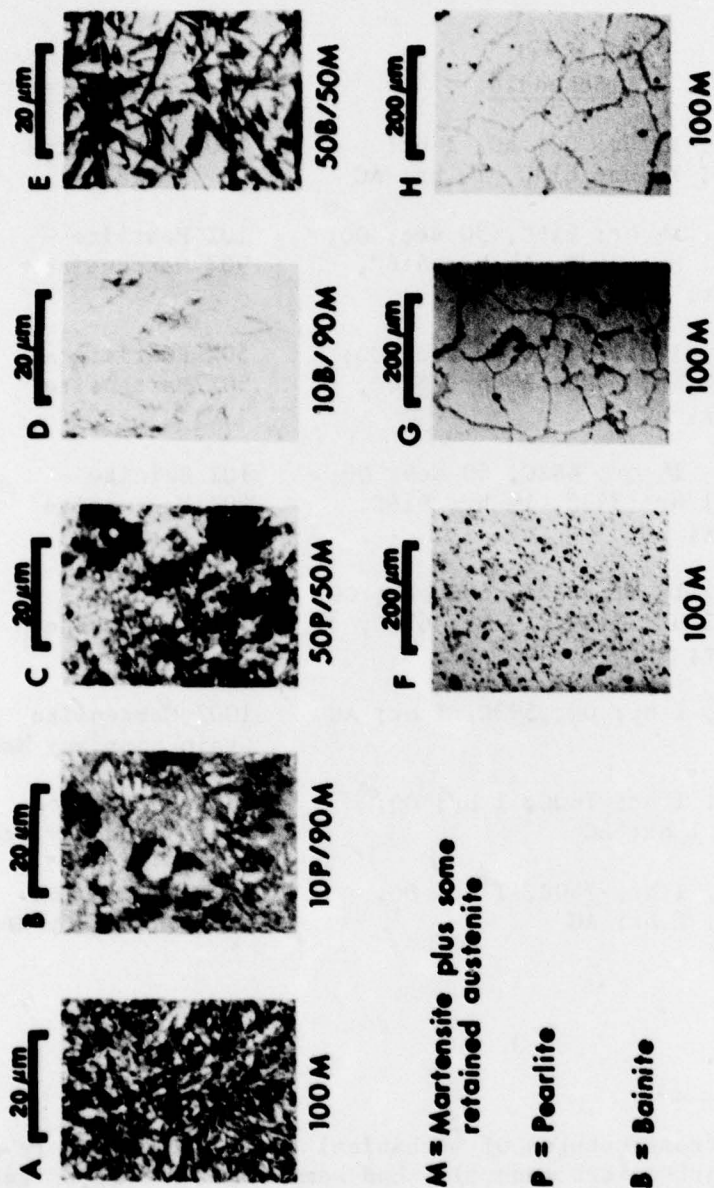
A 100M

B 10P/90M

C 50P/50M

D 10B/90M

E 50B/50M

F 100M

G 100M

H 100M

M = Martensite plus some retained austenite

P = Pearlite

B = Bainite

FIGURE 2. TYPICAL MICROSTRUCTURES OF ISOTHERMAL TRANSFORMATION PRODUCTS
(See Text for Details)

6

TABLE II. MECHANICAL PROPERTIES

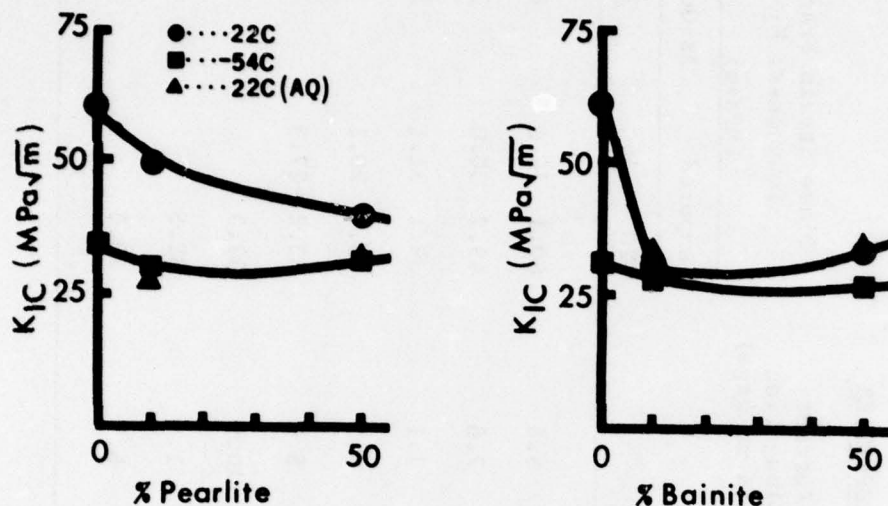| Material Condition (See Fig.2)	0.2% Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Percent Elongation (25.4 mm gage)	Plane Strain Fracture Toughness, K_{IC} (MPa√m)		
				Tempered		As-Quenched
				22C	-54C	22C
A	905	1122	8.8	60.1	34.6	(a)
B	804	1198	7.8	49.7	30.0	25.7
C	711	1203	7.1	39.1	31.1	32.0
D	858	1162	9.2	30.1	28.1	34.1
E	844	1140	8.9	33.2	27.3	34.7
F	949	1214	10.9	37.5		
G	1233	1316	1.3	32.5		
H	760	1038	6.7	40.3		

(a) Specimens broke during fatigue pre-cracking.

7

FIGURE 3.   INFLUENCE OF NON-MARTENSITIC ISOTHERMALLY TRANSFORMED PHASES
ON PLANE-STRAIN FRACTURE TOUGHNESS (K_{IC}).

In particular, compare materials A and F which both result in quenched-
and-tempered martensitic structures having similar tensile properties
and grain sizes. The presence of the grain boundary carbide network
causes almost a 40% reduction in plane-strain fracture toughness.

CONCLUSIONS

The presence of non-martensitic isothermal transformation products
seriously degrades the fracture resistance of hypereutectoid silico-
manganese steel. The effect is moderate for pearlite but catastrophic
for bainite. The steady decrease in fracture toughness with increasing
pearlite supports the idea that cracking proceeds by fracture of pearl-
ite colonies rather than along intergranular films of cementite. On
the other hand, the precipitous drop in fracture toughness with only
modest amounts of bainite is more consistent with the notion of frac-
ture along carbide films formed along both prior austenite and bainitic
ferrite grain boundaries. The deliberate introduction of continuous
carbide networks in the grain boundaries also seriously degrades frac-
ture toughness. No correlations between strength, ductility, and frac-
ture toughness exist in the material conditions evaluated. Therefore,

8

an effective fracture control plan for engineering applications utilizing such materials necessitates both the maintenance of precise heat treatment schedules and the incorporation of microstructural surveys in the quality assurance scheme.

ACKNOWLEDGEMENTS

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